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**DEFENCE SCIENCE AND TECHNOLOGY ORGANISATION**

**AERONAUTICAL RESEARCH LABORATORY**

MELBOURNE, VICTORIA

Aircraft Materials Technical Memorandum 405

**FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO T56  
TURBINE BLADES AFTER 500 HOURS OF ENGINE USE**

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by

S.G. RUSSO

**91-16316**



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JUNE 1991

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**AIRCRAFT MATERIALS TECHNICAL MEMORANDUM 405, JUNE 1991.  
"FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO  
T56 TURBINES AFTER 500 HOURS ENGINE USE"  
S.G. RUSSO**

**Erratum:**

*"500 hour engine use" stated in this document is incorrect.  
These differently coated turbine-blades have in fact seen 1500 hours service in a T56 engine.*

The title and further reference in the text should therefore read **1500 hours** and not **500 hours**.

**AIRCRAFT MATERIALS TECHNICAL MEMORANDUM 405, JUNE 1991.  
"FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO  
T56 TURBINES AFTER 1500 HOURS ENGINE USE"  
S.G. RUSSO**

This error arose when maintenance technicians overhauling this particular engine [No AE 107087] communicated that the operational service interval (since last engine overhaul) of 500 hours coincided with the total engine-life seen by this particular batch of blades. We have now discovered that these blades were first installed in this engine in 1983 and in 1986 had already seen 1000 hours of engine operation.

A large number of T56 blades coated with six different modifications of aluminide-coating were introduced into this T56 engine in 1983 as part of a long-term "Rainbow-trial" initiated by Officers of the DSTO Materials Research Laboratory. The first two stages of coating performance evaluation at 500 and 1000 hours has been reported in a paper authored by :- J. Cocking, P. Richards and G. Johnston; "Comparative Durability Of Six Coating Systems On First-stage Gas Turbine Blades In The Engines of a Long-Range Maritime Patrol Aircraft", Surface and Coating Technology, 36, [1988], 37-47. It is regretted that particular reference to this document was not previously cited in Tech Memo 405.

*Each place in the text where 500 hour life is noted should be corrected to now read 1500 hours. Please make this correction to the front cover, summary page and Document Control Data Sheet as well.*

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Aircraft Materials Technical Memorandum 405

**FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO T56  
TURBINE BLADES AFTER 500 HOURS OF ENGINE USE**

by

S.G. RUSSO

**SUMMARY**

*The evaluation of several protective coatings that have been applied to IN738-LC first stage turbine blades in two military gas turbines for an interval of 500 flight hours indicated variations in the extent of hot-corrosion resistance. The coatings included two conventional nickel aluminides, an aluminide modified with chromium, a platinum-modified aluminide, an aluminide modified with silicon and a platinum/rhodium-modified aluminide. The aim of this trial is to find a coating that will withstand up to 3000 engine operating hours. Detailed metallographic examinations as well as scanning-electron-microscopy were employed in the evaluation. Results indicated that the precious metal aluminides offered the best protection with the chromium-aluminide being the most susceptible to hot-corrosion. The conflicting results of the basic aluminide coatings reaffirms the point that no firm conclusions can be made solely on one assessment. Furthermore, there was no apparent cracking in any of the coatings.*



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## 1. INTRODUCTION

Conventional turbine blades fitted to aero-engines are manufactured almost exclusively from nickel or cobalt based superalloys. However, gas turbine technology has recently reached an era where these superalloys can no longer withstand the extreme environmental conditions they frequently experience. The metal temperatures at which these blades are required to operate are approaching a level of significantly reduced lifetime, due to inherently increased oxidation and hot corrosion rates. A reasonable approach to overcoming - or at least minimising these difficulties - is the application of protective coatings on the superalloy. Aluminide and modified aluminide coatings are presently the most popular coatings in use.

Protective metallic coatings can markedly improve the resistance of superalloys against high and low temperature hot-corrosion. The added protection is achieved through the formation of protective oxide barriers (usually  $\alpha$ -Al<sub>2</sub>O<sub>3</sub> and Cr<sub>2</sub>O<sub>3</sub>) which inhibit the subsequent reactions between the coating and the environment. It has been shown that the  $\alpha$ -Al<sub>2</sub>O<sub>3</sub> barrier layer is the most effective in reducing high temperature oxidation <sup>(1)</sup>.

A detailed investigation is currently being made on the performance of six different aluminide-based coatings. They were applied to the first-stage turbine blades in the engines of two long range maritime patrol aircraft operated by the Royal Australian Air Force (RAAF). Trials were conducted using two aircraft flying similar missions thus allowing comparisons to be made.

The RAAF anticipated the investigation would be carried out from 500 hours to 3000 hours of flight time at intervals of no more than 500 hours, with the hope of at least one aluminide-based coating surviving the full 3000 flight hours.

This paper investigates the performance of the six different coatings after 500 hours using both optical and scanning-electron-microscopy. A paper detailing the qualitative results from the Energy Dispersive Spectroscopy (EDS) of the coatings will follow shortly.

## 2. TYPES OF COATINGS INVESTIGATED

A total of twelve turbine blades were received from the RAAF Allison T56-A-14 engine (Serial No. AE 107087) after having experienced 500 hours of engine use.

Prior to service, each blade had been coated to withstand the effects of hot-corrosion by one of six different coatings, resulting in six

pairs of similarly coated blades. The coated blades were positioned around the turbine disc as illustrated in Figure 1. It is important to emphasise the fact that the similarly coated blades investigated in this report were diametrically opposed to one another on the turbine-disc and thus separated by approximately 180 degrees.

The six coatings applied to the vacuum cast nickel-base superalloy, designated IN-738LC whose weight composition is shown in Table 1, are as follows : (i) two nickel aluminides designated A and B; (ii) a platinum-modified aluminide; (iii) an aluminide modified with platinum/rhodium; (iv) a chromium-modified aluminide; and (v) an aluminide modified with a silicon slurry.

### 3. BLADE SURFACE FEATURES

On receiving the blades a visual and optical inspection using a stereo-microscope was initiated. Both the convex and concave sections of the blades were examined. Determination of the degree of spallation was enhanced using dye penetrants.

The concave section of each blade demonstrated some degree of discolouration and spallation. Dark regions of localised attack on the two aluminides and chromium-modified aluminide together with areas of pitting and surface roughening gave evidence of the initial signs of hot-corrosion. Spallation was most evident on both aluminide samples. One encouraging point was the absence of cracking in either the coating or root platform.

Discolouration of the convex sections to a 'rainbow' of colours ranging from dark brown to bright orange was clearly a major feature. Although spallation attack was evident on the convex sections, it was minimal when compared to the attack on the concave sides.

### 4. METALLOGRAPHIC EXAMINATION OF COATINGS

A comparative evaluation of the coated blades could only be achieved through careful preparation. It was also essential that the protective oxide layer ( $\text{Al}_2\text{O}_3$ ,  $\text{Cr}_2\text{O}_3$  etc.) be kept intact. Furthermore, it has been well documented that the likelihood of forming soluble compounds on the coating is relatively high. These compounds may have a detrimental effect on the protection of the blade. For this reason kerosene and ethanol, instead of water, were used in the metallographic preparation.

Once the sections of blade material had been meticulously prepared, a range of metallographic analyses, including both optical and electron microscopy, were performed.

#### **4.1 Blade Preparation Procedure**

Prior to sectioning, each blade was coated with an epoxy resin, vacuum-impregnated and dried at 30° C overnight. This reduced the likelihood of the oxide being removed during polishing/sectioning.

Another problem confronted was the "rounding of edges" of the sample. This was caused by the difference in hardness between the oxide/mounting material and the superalloy. To overcome this problem a layer of gold was sputtered onto the epoxy which enabled a nickel-plate to be deposited on the surface. The nickel-plate, being hard and resistant to most forms of wear, aided in keeping the samples' edges level. A schematic cross-section of the blade after this preparation is illustrated in Figure 2.

Blades were sectioned transversely through the aerofoils near the root platform, at the mid-span and near the shroud tip (Figure 3). Since there was a lack of unused coated blades, the initial condition of the coating was unknown. It was therefore essential to have a typical coating prior to use for each coating system that could be used for comparative purposes. These were obtained from areas on the shrouds which were assumed to be equivalent to an unused coating. This assumption can be justified when one considers the shroud is located outside the main air-stream. Sectioning the shroud parallel to the blade's vertical axis, as shown in Figure 3, resulted in micrographs which were used to determine the degree of attack a blade had suffered.

A total of five samples from each blade were sectioned, totalling sixty samples altogether. Thereafter, the samples were mounted in a low shrinkage epoxy resin and polished to a one micron finish for metallographic examination.

#### **4.2 Shroud Platform Coating Microscopy**

The microstructures of the shroud coatings varied considerably as illustrated in Figure 4. It should be stressed that etchants were not used in order to keep the coatings in their initial states. Only a brief description of the processing procedure is given as full details for each coating was not available. In most cases this information is proprietary.

##### **4.2.1 Nickel Aluminide coatings**

This was a conventional diffusion-aluminide coating applied by the pack-cementation process and was supplied by two vendors differentiated by naming them A and B. The coating thickness along the chord lengths of both types was between 80-100 microns, with thicker sections at the leading and trailing edges.



Microstructural studies of the coating indicated that both samples had similar inward-diffusion, high-activity coatings with three distinct regions. The outer-zone consisted of a large concentration of fine precipitates that appeared to be more abundant in the Aluminide A sample. Below this a precipitate-free region was evident leading to the inner diffusion layer. The diffusion layer had formed when the nickel, originally in this region of the superalloy, diffused outward to interact with the inwardly diffusing aluminium from the pack.

#### 4.2.2 *Platinum-modified aluminide*

It was initially reported <sup>(2)</sup> that the addition of one or more noble metals such as platinum or rhodium to an aluminide coating would substantially increase its lifetime in a sulfidizing environment by a factor of two to three.

The platinum-modified aluminide coating had been applied by firstly electroplating a thin platinum layer (several microns thick) on a substrate followed by pack-aluminising. The coating had a thickness of 60-70 microns in the alloy chord lengths.

The microstructure from Figure 4 illustrates a bright area near the coating's edge. This single-phase structure, being brighter than the rest of the coating, must be rich in heavy elements. The heavy elements may be platinum, tungsten or tantalum etc. Therefore, the coating shows all the typical features of an outward-diffusing, low-activity aluminide coating with an additional (Pt, W or Ta etc.) rich layer on top of the coating.

#### 4.2.3 *Platinum and Rhodium-modified aluminide*

The process used to deposit the Pt/Rh coating was very similar to that of the platinum-modified aluminide. The process apparently produces a Pt-rich outer layer and a Rh-rich inner layer adjacent to the substrate material. Thicknesses of the coating varied between 80 microns at the chord lengths and up to 100 microns on the leading and trailing edges making it thicker than the platinum aluminide sample.

The coating also exhibited an outward-diffusing, low-activity microstructure similar to the platinum-modified aluminide. It should be noted that dark precipitates near the edge and within the coating, most probably  $\alpha\text{-Al}_2\text{O}_3$ , were a predominant feature of these two coatings. The major difference between this coating and the previous one was a higher distribution of fine precipitates in the middle of the coating resulting in a more complex microstructure.

#### 4.2.4 *Chromium-modified aluminide*

The role of chromium in hot-corrosion of superalloys has been well documented and the beneficial effect of increasing its concentration to inhibit sulfur induced degradation is well known. Chromium has been found to improve the hot-corrosion resistance of nickel aluminide

coatings in two ways. Firstly, by suppressing the formation of  $\gamma'$ -Ni<sub>3</sub>Al which is particularly susceptible to hot-corrosion and secondly, by promoting the stable  $\alpha$ -Cr phase <sup>(3)</sup>.

The chromium present in this coating had been electrophoretically deposited on the substrate prior to heat-treatment resulting in an outward-diffusing, low-activity coating. The coating thickness varied with some areas on the chord being 60 microns thick and other areas being up to 70 microns thick.

As Figure 4 illustrates, the coating consisted of a mixture of dark and bright precipitates in both the single outer-zone and the diffusion layer. Compounds in the diffusion-zone were somewhat coarser than the other coatings and, by comparison, resulted in a thicker zone.

#### 4.2.5 Silicon-modified aluminide

Silicide diffusion coatings for gas turbine applications have been well documented since 1960 <sup>(4)</sup>. The formation of an amorphous SiO<sub>2</sub> subscale has been identified as the source of its excellent hot-corrosion resistance.

The high-activity silicon-modified aluminide coating had been formed by depositing a slurry of aluminium and silicon powder in an organic binder followed by a diffusion heat-treatment. It was the thinnest of the coatings examined, being only 50 microns thick in areas along the chord, but reported by numerous authors to be brittle <sup>(4)</sup>.

Segregation of large multi-phase precipitates on the grain-boundaries penetrating the coating edge to the substrate were evident. Furthermore, fine precipitates in the outer and diffusion-zones distinguished this coating from the others.

#### 4.3 Microscopy of Coatings after 500 hours of engine use

As previously mentioned, all blades were sectioned to give five samples for each specimen (see Figure 2). Comparisons of the coatings with the shroud microstructures were made using areas of the blade that had experienced the most severe attack. This occurred at the mid-section of the concave leading edge of the blade due to its high heat-transfer coefficient. The micrographs illustrated in Figure 5 are taken from this region.

The microstructures of the coatings after 500 hours of engine use (Figures 5, 6 and 7) clearly illustrate areas of attack and/or microstructural changes in all six coatings. The platinum, platinum/rhodium and silicon-modified coated blades exhibited only slight hot-corrosion attack while the two aluminide and chromium-modified aluminide coatings showed regions of attack ranging from superficial to severe.

#### 4.3.1 *Nickel Aluminide Coatings*

Large discrepancies regarding the extent of attack occurred in the aluminide coatings of A and B. Blade No. 8 and No. 59, both coated with aluminide A, exhibited minimal and severe hot-corrosion attack respectively. These discrepancies can be more easily understood when one realises the blades were diametrically opposed to one another on the turbine-disc and although the gas-phase environment would have been similar for both blades the solid-phase environment may have differed. At this stage no reasonable explanation can be given for these conflicting results. Hopefully, qualitative analysis of the microstructure included in the next report will give some insight.

Similarly, the microstructure of coating B deposited on blade 29, apart from demonstrating precipitate coarsening, showed little if any form of attack. Blade 79 which was also coated with coating B, displayed severe degradation of the coating penetrating as deep as the diffusion layer. Areas of attack for both these blades (Figure 6 (a) and (b)) can be compared with micrographs (a) and (b) in Figure 5 respectively.

#### 4.3.2 *Platinum-modified aluminide*

Microstructural features of the platinum-coated blade after 500 hours indicated no apparent platinum-rich layer. The bright platinum-rich layer that appeared on the shroud section was absent. The platinum is believed to have dispersed throughout the coating after prolonged use at high temperatures. Another feature of the coating was the coarsening of the precipitates in the diffusion-zone in addition to grain-boundary precipitation. Overall, the thickness of the coating remained relatively consistent with that of the shroud.

#### 4.3.3 *Platinum and Rhodium-modified aluminide*

The bright outer region, initially present in the shroud microstructure, was absent. This may indicate the dispersion of platinum and other heavy elements throughout the coating. Despite this, the coating demonstrated a good resistance to attack. The increase in abundance of dark precipitates, particularly near the coating's edge, was caused by the outward diffusion of nickel from the substrate. Diffusion of compounds within the diffusion-zone towards the coating was also apparent. Close examination of the microstructure revealed a reduction in the number of precipitates in the coating compared to the shroud, caused by coalescing in the NiAl compound at high temperatures.

#### 4.3.4 *Chromium-modified aluminide*

Investigations of the six coatings indicated the chromium-modified aluminide experienced the most degradation. Close to the leading edge, where corrosion was the most severe, the coating was frequently found to be completely destroyed, allowing direct attack of the metal substrate. Moreover, it was impossible to distinguish the location of the diffusion zone. Finger-like penetrations due to preferential attack of substrate carbides were present in the substrate (see Figure 6 (e)). A

corrosion-scale had also formed on the blade. Despite these problems, the attack appeared to be localised solely on the concave leading edge with only minimal penetration on the convex side.

#### 4.3.5 *Silicon-modified aluminide*

A considerable change in microstructure appeared in the silicon-modified aluminide coating after having experienced 500 flight hours. The coating was thinner (only up to 40 microns thick) with small precipitates, initially present in the shroud, coalescing to form coarser precipitates. Grain-boundary precipitation was evident and a large needle-like structure beneath the diffusion-zone (most probably the sigma phase) was visible. Despite these problems the coating remained relatively intact with little evidence of hot-corrosion.

## 5. DISCUSSION

The prime objective of this coating evaluation programme was to compare the hot-corrosion resistance of various coatings and thus to be able to rank them. Ranking the coatings based on their overall corrosion-resistance gave the results shown in Table 2. The ranking was determined by taking into account the depth of attack and whether it penetrated to the substrate, the number of different sites at which the coating had been attacked, the coating's viscosity (ie. whether the depleted coating had formed a continuous or discontinuous layer), and the coarsening of precipitates in the coating.

A similarity of all coatings studied was their method of protection with each one preventing attack of the underlying substrate by forming a tenacious, continuous oxide-scale. The composition of the scale product influenced its corrosion-resistance and had the added benefit that it reduced ion migration.

Undoubtedly, the platinum and platinum/rhodium-modified aluminide coatings proved to be the best, based on their excellent hot-corrosion resistance. There was no evidence of attack either at the leading or trailing edges of the blades. Presumably, they owe their superior resistance to the presence of a  $PtAl_2$  phase near the blades surface (5).

The use of platinum in such coatings has its origins in the belief that it would act as a diffusion-barrier system that would overcome the problem of aluminium migration to the substrate. The result would be a high activity of aluminium at the surface, where it was essentially required. It is now thought that the platinum acts as a diffusion medium for the aluminium rather than a diffusion-barrier resulting in a simultaneous formation of a nickel-aluminide subsurface structure in addition to the generation of a platinum-aluminium intermetallic skin (6). The micrographs taken after 500 hours and the absence of the bright

region, indicated that the  $\text{PtAl}_2$  phase had dispersed to form a solid solution.

Results for the silicon-modified aluminide coating indicated favourable results in combatting the effects of hot-corrosion with the formation of a  $\text{SiO}_2$  scale, in addition to the  $\text{Al}_2\text{O}_3$  and  $\text{Cr}_2\text{O}_3$  scales. The coating contained no evidence of cracking, which was highly unexpected since the ductile/brittle transition temperature (DBTT) of this coating was much higher than the others <sup>(7)</sup>. Application of the aluminium and silicon slurry was carried out manually and is thus dependent on the skill of the operator to achieve a coating with consistent thickness. Figure 7 illustrates areas of local thinning and thickening in the coating brought about by the blades complex shape.

The conflicting results of attack obtained in the aluminide coatings can only be fully explained by detailed analyses that are to be undertaken for the next report. An educated guess to this anomaly may focus on the fact that one blade experienced accelerated oxidation associated with the deposition of a higher concentration of sulfur and alkali salts, the main causes of hot-corrosion. Sodium compounds, present in sea salt spray, may have been ingested into the main air-stream thereby bringing it in contact with the stationary guide vanes. Considering the continual build-up of salt and the frequent passage of gas through the vanes, a stage would have been reached where the salt could no longer remain deposited on the vane. The result of this may have been the breaking away of salt and its bombardment with the rotating turbine blades. This phenomenon is commonly termed 'salt shedding' <sup>(8)</sup> and may explain the fact that some blades exhibited a more severe form of degradation than others. The salt, now present on particular blades, and sulfur occurring in the fuel may have reacted to form the destructive  $\text{Na}_2\text{SO}_4$  compound, condensing as a molten layer on the metal. The major reaction product is a non-protective oxide layer, often containing fragments of metal; the alloy/scale interface is very irregular and in the alloy immediately beneath the interface small amounts of sulphides, usually chromium-rich, are commonly present <sup>(9)</sup>. Alternatively, the excellent condition of the other aluminide blades may be caused by the absence of these sulfur-containing compounds.

Finally, an explanation to the unexpected degradation of the chromium-modified aluminide coating, as with the aluminide coatings, can only be given with more detailed analysis. Hopefully this will give a better insight into the various mechanisms taking place in the six coatings. One reason may be that  $\text{Cr}_2\text{O}_3$  is not as protective as  $\text{Al}_2\text{O}_3$  for oxidation resistance, since its growth rate is greater than that of  $\text{Al}_2\text{O}_3$  and, all other things being equal, it is preferred that the protective oxide increases in thickness at the lowest possible rate <sup>(10)</sup>. Nevertheless, the entire growth zone, including the sigma-phase, appears to have been totally consumed.

## 6. CONCLUSIONS

It should be emphasised that none of the coatings were totally resistant to attack, since even the most protective coating had degraded in localised areas. On the basis of the metallographic examination conducted thus far the following remarks can be given:

- (1) Initially, each coating showed a distinctive microstructure consisting of an outer-zone, an inner-zone and a diffusion layer.
- (2) The most severe form of attack, characteristic of hot-corrosion, was located at the middle of the blade near the concave leading edge.
- (3) After 500 hours of engine operation, the condition of the first stage high-pressure blades with the six different coatings ranked from best to worst was platinum/rhodium-modified aluminide > platinum-modified aluminide > silicon-modified aluminide > conventional nickel aluminide > chromium-modified aluminide. The mechanisms of degradation of these coatings will be the subject of a separate report.
- (4) An anomaly existed in the results for the conventional aluminide coatings A and B probably caused by being exposed to different conditions while in the same engine. From this it should be stated that until further studies are carried out, including burner-rig testing, no clear conclusions can be given.
- (5) There was no evidence of cracking in any of the six coatings.
- (6) Each blade demonstrated some degree of spallation and discolouration which can only be described as superficial.

## 7. ACKNOWLEDGEMENTS

The metallurgical preparation side of this report was carried out by Mr. P Rohan and the author. The author would also like to thank Dr. G. McAdam for his invaluable advice and discussions. This work was supported by the R.A.A.F., Task No. 88/025, Dr. Neil Ryan.

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Table 1

Chemical Composition of IN-738LC (\*)

Element	Concentration (wt. %)
Nickel (Ni)	balance
Chromium (Cr)	16.0
Cobalt (Co)	8.5
Aluminium (Al)	3.4
Titanium (Ti)	3.4
Tungsten (W)	2.6
Molybdenum (Mo)	1.75
Tantalum (Ta)	1.75
Niobium (Nb)	0.9
Iron (Fe)	0.5 (a)
Silicon (Si)	0.3 (a)
Manganese (Mn)	0.2 (a)
Carbon (C)	0.11
Zirconium (Zr)	0.06
Boron (B)	0.01

(\*) IN-738LC is a specification of the Inconel Alloy Group.

(a) Maximum composition.



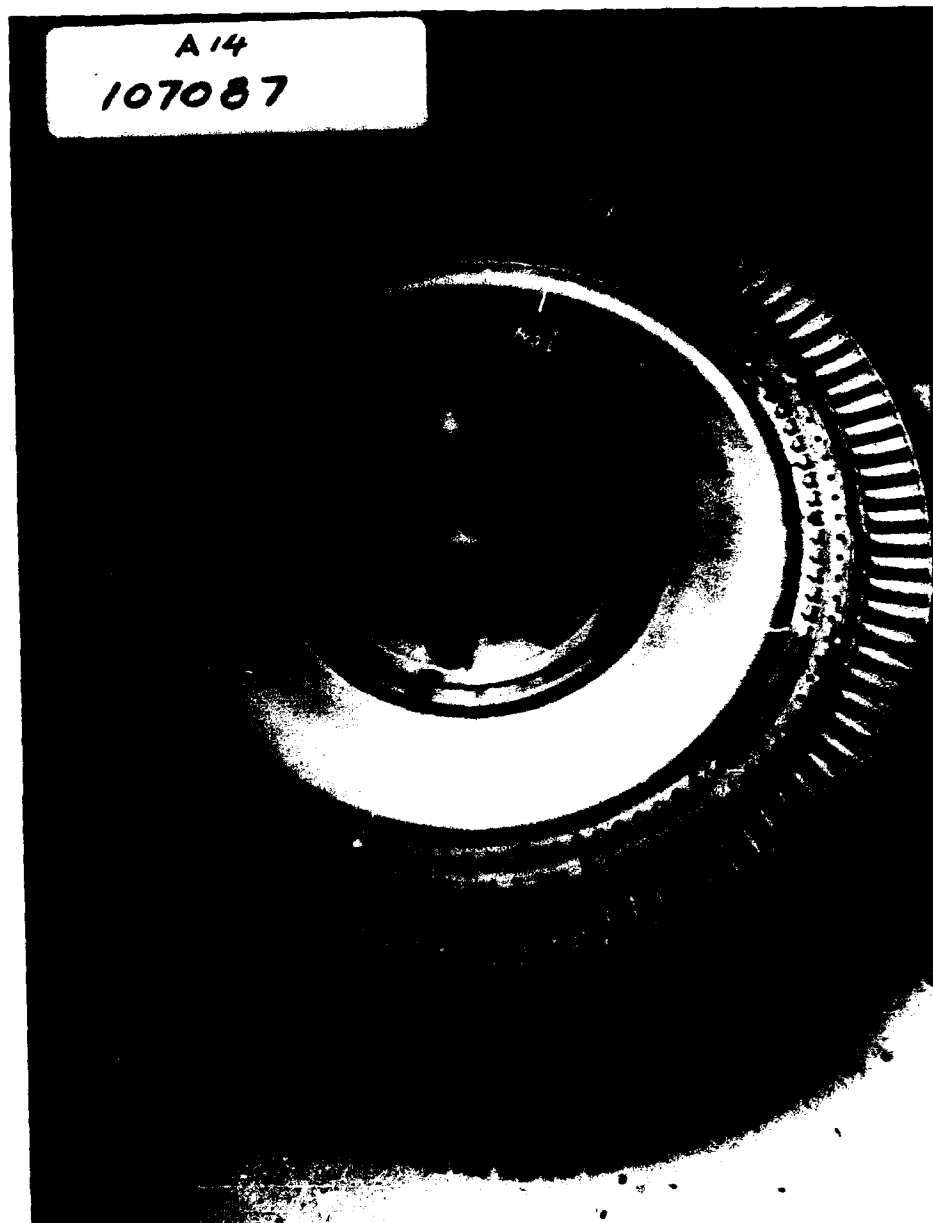
**Table 2**

**Ranking of Coatings (b)**

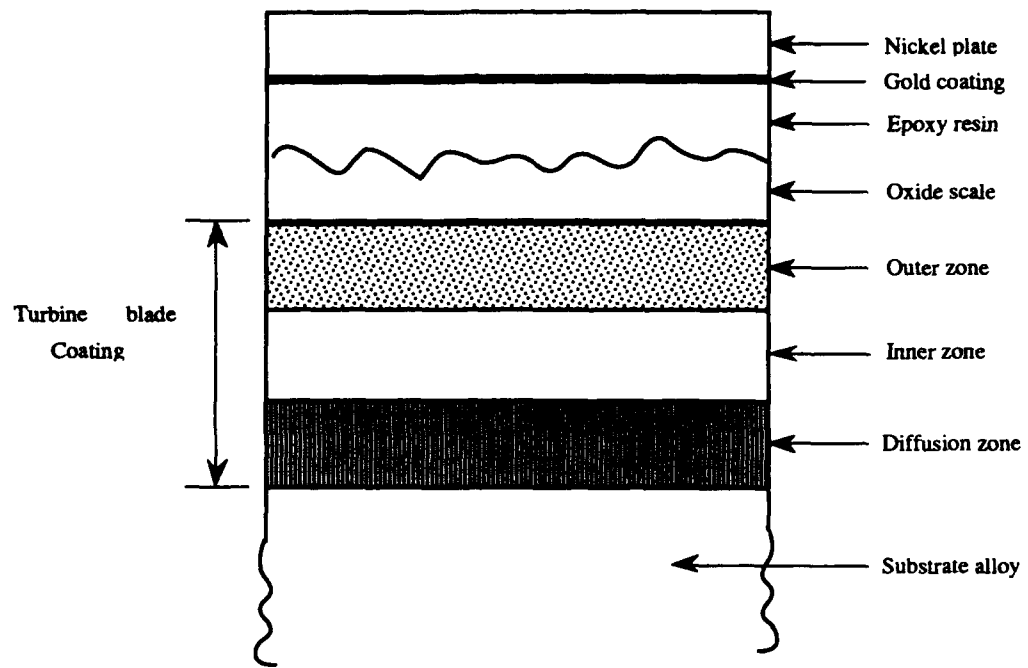
Type of coating	Av. initial coating thickness ( $\mu\text{m}$ )	Av. final coating thickness ( $\mu\text{m}$ )	Hot-corrosion resistance	Ranking
Ni aluminide A	85	80	very good to poor	4 *
Ni aluminide B	80	55	very good to poor	5 *
Pt-modified aluminide	70	65	excellent	2
Pt/Rh-modified aluminide	85	85	excellent	1
Si-modified aluminide	50	40	good	3
Cr-modified aluminide	65	0	very poor	6

\* variations in results as illustrated in Figure 5

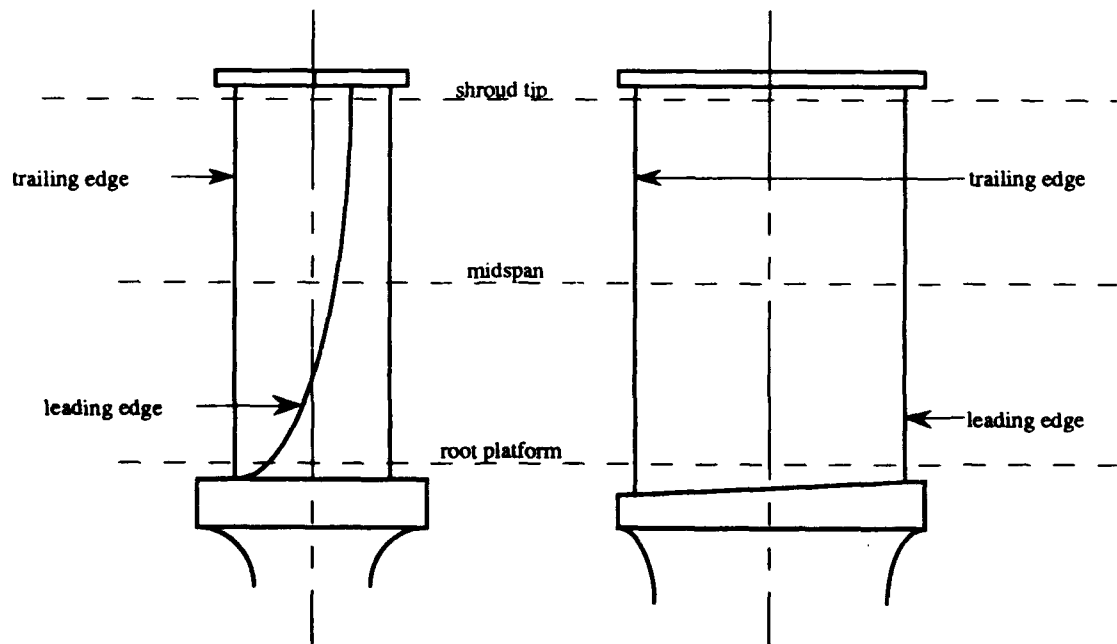
(b) all values quoted in the above table were obtained from areas of maximum attack, namely the mid-section of the concave leading edge.



**Figure 1** First stage turbine wheel prior to disassembly illustrating the relationship between a blade's designated number and its polar coordinates.



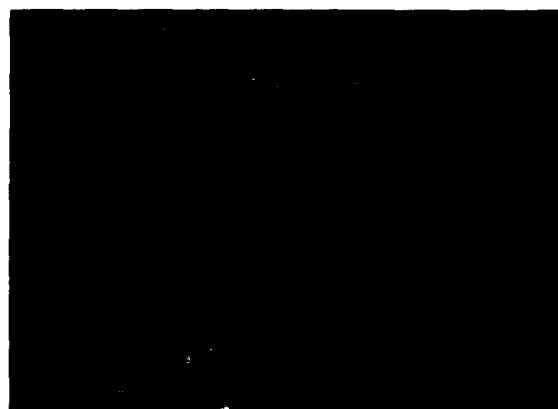
**Figure 2** Cross-section of the blade after preparation.



**Figure 3** Schematic diagram of a turbine blade illustrating the positions where they were sectioned.



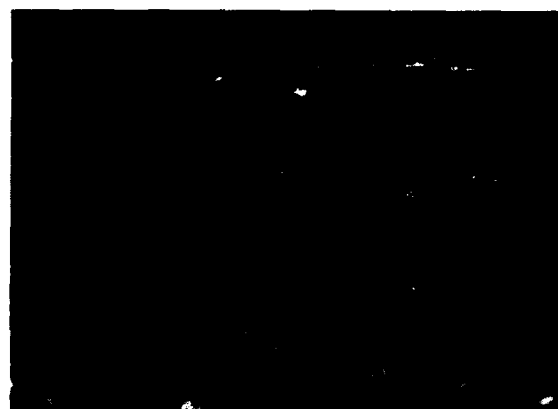
(a) Ni aluminide A



(b) Ni aluminide B



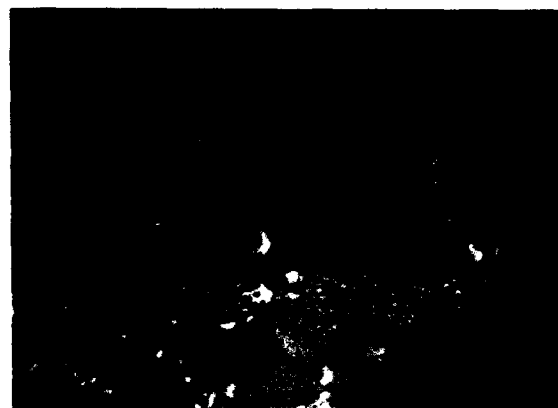
(c) Pt-modified aluminide



(d) Pt/Rh-modified aluminide



(e) Cr-modified aluminide



(f) Si-modified aluminide

20  $\mu\text{m}$

**Figure 4** Scanning electron micrographs of the six shroud coatings (unetched).  
All microraphs were taken parallel to the vertical axis of the blade  
and have identical magnifications as shown by the micron-marker.



(a) Ni aluminide A



(b) Ni aluminide B



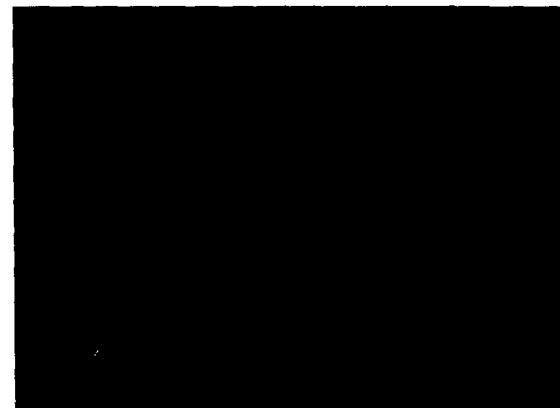
(c) Pt-modified aluminide



(d) Pt/Rh-modified aluminide



(e) Cr-modified aluminide

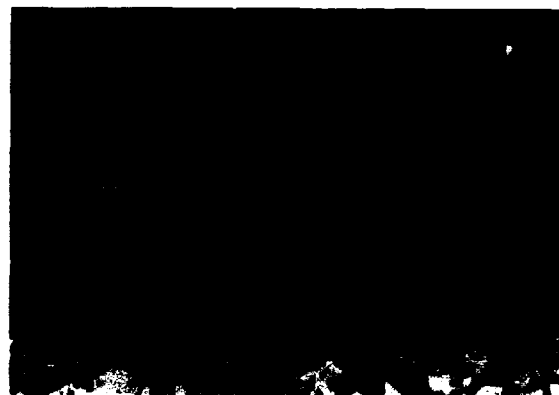


(f) Si-modified aluminide 30  $\mu\text{m}$

**Figure 5** Scanning electron micrographs of the six shroud coatings at the middle of the concave leading edge after 500 hours of engine use (unetched). Magnifications were equivalent to the micron-marker shown in (f).



(a) Nickel aluminide A



(b) Nickel aluminide B 30  $\mu$ m

**Figure 6** Micrographs of the two aluminide coatings illustrating the extreme degradation on the mid-section of the concave leading edges.



(a) Si-modified aluminide



(b) Pt-modified aluminide 200  $\mu$ m

**Figure 7** Comparing the trailing edge micrographs of the (a) silicon-modified aluminide with the (b) platinum-modified aluminide. The latter is the more desirable coating since the silicide coating shows large variations in thickness.

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4. TITLE <b>FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO T56 TURBINE BLADES AFTER 500 HOURS OF ENGINE USE</b>		5. SECURITY CLASSIFICATION (PLACE APPROPRIATE CLASSIFICATION IN BOXES) (E. SECRET (S), CONF. (C) RESTRICTED (R), UNCLASSIFIED (U)).  <div style="display: flex; justify-content: space-around;"> <div style="border: 1px solid black; padding: 2px; text-align: center;">U DOCUMENT</div> <div style="border: 1px solid black; padding: 2px; text-align: center;">U TITLE</div> <div style="border: 1px solid black; padding: 2px; text-align: center;">U ABSTRACT</div> </div>	6. NO. PAGES  <b>19</b>	7. NO. REFS.  <b>10</b>
8. AUTHOR(S) <b>S.G. RUSSO</b>		9. DOWNGRADING/DELIMITING INSTRUCTIONS  <b>Not applicable</b>		
10. CORPORATE AUTHOR AND ADDRESS <b>AERONAUTICAL RESEARCH LABORATORY 506 LORIMER STREET FISHERMENS BEND VIC 3207</b>		11. OFFICE/POSITION RESPONSIBLE FOR: <b>RAAF</b> SPONSOR _____ SECURITY _____ DOWNGRADING _____ APPROVAL <b>CAMD</b>		
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14. DESCRIPTORS <b>Allison T56 engines Hot-corrosion Turbine blades Protective coatings Heat resistant alloys</b>			15. DISCAT SUBJECT CATEGORIES <b>2105 1103 130805</b>	
16. ABSTRACT <i>The evaluation of several protective coatings that have been applied to IN738-LC first stage turbine blades in two military gas turbines for an interval of 500 flight hours indicated variations in the extent of hot-corrosion resistance. The coatings included two conventional nickel aluminides, an aluminide modified with chromium, a platinum-modified aluminide, an aluminide modified with silicon and a platinum/rhodium-modified aluminide. The aim of this trial is to find a coating that will withstand up to 3000 engine operating hours. Detailed metallographic examinations as well as scanning-electron-microscopy were employed in the evaluation. Results indicated that the precious metal aluminides offered the best protection with the chromium-aluminide being the most susceptible to hot-corrosion. The conflicting results of the basic aluminide coatings reaffirms the point that no firm conclusions can be made solely on one assessment. Furthermore, there was no apparent cracking in any of the coatings.</i>				

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16. ABSTRACT (CONT).

17. IMPRINT

**AERONAUTICAL RESEARCH LABORATORY, MELBOURNE**

18. DOCUMENT SERIES AND NUMBER

Aircraft Materials Technical  
Memorandum 405

19. COST CODE

33 3010

20. TYPE OF REPORT AND PERIOD COVERED

21. COMPUTER PROGRAMS USED

22. ESTABLISHMENT FILE REF.(S)

23. ADDITIONAL INFORMATION (AS REQUIRED)